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AGARD'S APPROACH TO FATIGUE AND FRACTURE RESISTANCE, (U)  
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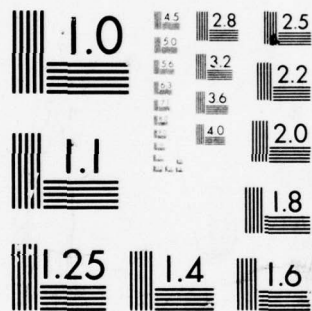
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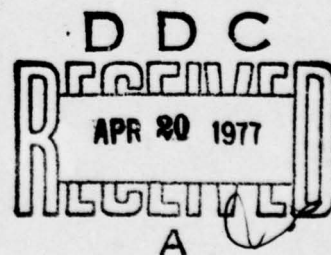
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**AGARD'S APPROACH TO FATIGUE AND FRACTURE RESISTANCE**

BY

R. J. H. WANHILL



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AGARD'S APPROACH TO FATIGUE AND FRACTURE RESISTANCE

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SUMMARY

AGARD has organised two recent meetings on alloy microstructure influences on fatigue and fracture, principally in high strength structural materials. The present paper briefly reviews the subject, covering fatigue strength, fatigue crack propagation, fracture toughness, and stress corrosion in aluminium alloys, titanium alloys, and high strength steels.

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## INTRODUCTION

At the invitation of the AGARD Structures and Materials Panel, a number of specialists presented pilot papers on fatigue and fracture in aerospace structural alloys at a meeting in Milan, April 1973<sup>1</sup>. Discussions in an Ad Hoc Group resulted in a recommendation to the Panel to organise a specialists meeting on Alloy Design for Fatigue and Fracture Resistance. This meeting was held in Brussels, April 1975.

Essentially, the AGARD approach has been to examine in detail the influence of microstructure on fatigue strength, fatigue crack propagation, fracture toughness and stress corrosion in three types of materials: aluminium alloys, titanium alloys, and high strength steels.

The present paper reviews these topics. During its preparation much of the original literature referenced by the specialists was consulted, as well as many other publications. This was done in the hope of avoiding excessive selectivity on my part. All credit is due to the specialists, whose comprehensive treatments made this review possible.

## FATIGUE STRENGTH

The high-cycle fatigue strengths of high strength materials are compared in Table I. The table shows that aluminium alloys and maraging steels have poorer (fatigue strength/ultimate strength) ratios than low alloy high strength steels, and that there is a wide spread in titanium alloy fatigue strength. Much work has been devoted to improving the fatigue strengths of aluminium alloys, and to obtaining consistently high fatigue strengths in titanium alloys.

Most high-cycle fatigue testing has been carried out in an air environment, for which mechanical and microstructural parameters are likely to be dominating. Consequently, the following discussion is based on testing in air. However, it must be noted that corrosive environments have a strong effect on the fatigue strengths of aluminium alloys and steels (but not titanium alloys), so that continuing investigations should take environmental influences into account.



### Aluminium alloys

In high strength aluminium alloys fatigue cracks often nucleate at large second phase particles ( $\sim 5 \mu\text{m}$ )<sup>4,5</sup>, and a reduction in volume fraction of second phases can increase the fatigue strength, figure 1a. The increase is not large, and other results even indicate no improvement<sup>7</sup>.

The effect of fine particles or dispersoids has been investigated<sup>8,9</sup>. Figure 1b shows that dispersoids ( $\sim 1 \mu\text{m}$ ) produced by Ni and Mn additions to a ternary Al-5Mg-4Zn alloy can significantly improve the fatigue strength. On the other hand, reducing the concentration of small inclusions ( $0.1\text{--}0.2 \mu\text{m}$ ) in 2024 decreased fatigue life, figure 1c. The beneficial influence of these small inclusions was ascribed to their prevention of slip concentration in narrow bands<sup>9</sup>.

Attempts have been made to raise the fatigue strength by thermomechanical treatment (TMT)<sup>10-12</sup>. Results indicate better unnotched fatigue properties, figure 1d, but no improvement in notched fatigue life<sup>11</sup>. The improvement shown in figure 1d was attributed to a high density of entangled dislocations introduced by cold work during TMT<sup>10</sup>. However, this attribution cannot be generalized, and other influences, such as elimination of weak precipitate-free zones, may be important for different alloys<sup>13,14</sup>. The lack of improvement in notched fatigue strength is unexplained at the present time.

Limited data<sup>5,15,16</sup> on powder metallurgy aluminium alloys with very fine grain sizes indicate that improvements in fatigue strength over conventional materials are obtained for both smooth and notched ( $K_t = 3$ ) specimens. The disadvantage of powder metallurgy products is the cost premium<sup>5</sup>.

### Titanium alloys

Several investigations<sup>17-22</sup> have shown that refinement of primary  $\alpha$  grain size by processing has a beneficial influence on the fatigue strength of titanium alloys. For two-phase,  $\alpha$ - $\beta$  titanium alloys it has also been found that reducing the length of  $\alpha/\beta$  interfaces by altering the phase morphology is beneficial<sup>23</sup>, and that heat treatment to produce different microstructures affects fatigue life<sup>19</sup>. Figure 2 shows that replacement of equiaxed primary  $\alpha$  by coarse acicular  $\alpha$ , accomplished by air cooling thick forgings of Ti-6Al-4V from the  $\beta$  field, proved to be deleterious; but refinement of the acicular  $\alpha$  by water quenching the same forgings from the  $\beta$  field was beneficial.

Preferred orientation (texture) can significantly affect fatigue properties. A study<sup>24</sup> of highly textured Ti-4Al-4V plate showed that alignment of the hexagonal  $\alpha$  phase c-axis with the loading direction resulted in much lower

fatigue strengths below  $10^7$  cycles than alignment of the a-axis with the loading direction, figure 3. Somewhat less extreme effects of texture have also been reported<sup>21,25</sup>. Apparently, the textural influence is related both to changes in crack initiation mechanisms<sup>24</sup> and to strain localization in favourably oriented grains<sup>21,25</sup>.

### Steels

The classic work of STULEN, CUMMINGS and SCHULTE<sup>26</sup> showed that cracks initiate at inclusions in 4340 steel, and that large inclusions provide nucleation sites for high-cycle fatigue. THORNTON<sup>27</sup> critically reviewed a number of studies attempting to relate inclusion content and size to fatigue strength in low alloy, high strength steels. He concluded that no general trend exists, although in some cases increasing inclusion content and size apparently decreased fatigue strength.

The low fatigue strength of maraging steels is due to extensive cyclic softening<sup>28</sup>. In aged (high strength) maraging steel the source of cyclic softening seems to be the multiplication and rearrangement into cell structures of geometrically necessary dislocations associated with the age-hardening precipitates<sup>28</sup>. Thermomechanical processing can raise the fatigue strength by 30 %<sup>29</sup>, apparently because of the development of a dislocation cell structure (thereby inhibiting cyclic softening) and a coarsening of precipitates. The latter tend to inhibit formation and spreading of deformation bands and promote the homogenization of deformation by cross-slip<sup>29</sup>.

### FATIGUE CRACK PROPAGATION

A schematic of the dependence of fatigue crack propagation rate on stress intensity factor range is given in figure 4. At very low stress intensities there are often indications of threshold values,  $\Delta K_{TH}$ , for crack growth. At high stress intensities the crack rate accelerates greatly as the maximum stress intensity approaches the fracture toughness,  $K_c$ .

For fail-safe design, crack rates above  $0.1 \mu\text{m}/\text{cycle}$  are usually of greatest concern<sup>30,31</sup>. At the low end of this regime the basic microstructure exerts the most influence on crack rates in mild or nominally inert environments. The fact that crack rates for widely different materials are normalized by plotting them against stress intensity factor/Young's modulus<sup>32-34</sup> indicates that the modulus is the most important mechanical parameter. Particle size and distribution significantly influence crack rates only above  $1 \mu\text{m}/\text{cycle}$ <sup>35,36</sup>.



The environment often has an overwhelming influence on crack propagation rates. Consequently, each of the following sub-sections will conclude with a short discussion of environmental effects.

#### Aluminium alloys

Figure 5 shows the widest spread in fatigue crack propagation resistance in air for 2024-T3 sheet from seven different manufacturers, and also compares the T8 and T3 conditions<sup>37</sup>. The differences are not large: crack rates in 2024-T3 (Heat E) are 1.5-2 times those for 2024-T3 (Heat F); crack rates in 2024-T8 (Heat E) are 2-3 times those for 2024-T3 (Heat E).

Figure 6 shows that growth rates in 7075-T6 are generally higher, by a factor 3, than in 2024-T3 alloy. It seems likely that this difference is mainly metallurgical rather than due to differing susceptibility to environment-enhanced cracking<sup>35</sup>.

Recent work shows that within each of the 2000 and 7000 alloy series there is little difference in crack propagation rates in air up to rates of about  $0.5 \mu\text{m}/\text{cycle}$ . In one case the concentration of small inclusions ( $0.1\text{--}0.2 \mu\text{m}$ ) in 2024 was reduced<sup>9</sup>. In another instance three 7075 alloys of high, medium and commercial purity with respect to inclusion content were compared<sup>36</sup>. Lastly, 7475 and 7050 alloys in various tempers have been compared with 7075 in T6 and T7351 tempers<sup>5,38,39</sup>.

Crack propagation rates above about  $1 \mu\text{m}/\text{cycle}$  can be doubled by the presence of many large inclusions<sup>35,36</sup>, since these are nucleation sites for large voids which significantly advance the crack front<sup>40,41</sup>.

Extensive testing<sup>42,43</sup> has shown that the effect of environment in the fail-safe regime and gust frequency range is at least as significant as variations in mechanical fatigue crack growth rates. Results on the new 7000 series alloys, 7475 and 7050, are summarised in figure 7: crack rates in 3.5 % aqueous NaCl were 2-3 times faster than in dry air.

#### Titanium alloys

Figure 8 summarises fracture behaviour during crack propagation in titanium alloys in mild or nominally inert gaseous environments, including normal air. At very low propagation rates ( $< 0.1 \mu\text{m}/\text{cycle}$ ) the fracture surface consists largely of cleavage facets<sup>45-47</sup>, and there is evidence that coarser grain sizes give slower crack propagation rates<sup>48</sup>. However, in the regime relevant to fail-safe design (crack rates  $\geq 0.1 \mu\text{m}/\text{cycle}$ ) the propagation rates are affected mainly by microstructural variations other than grain size.

Figure 9 shows the influence of microstructure on crack growth rates in Ti-6Al-4V: there is a wide spread, often more than an order of magnitude. The lowest growth rates occurred in  $\beta$  annealed material, a result which appears to be general<sup>38,47</sup>. PATON et al<sup>47</sup> found that secondary cracking, at  $\alpha/\beta$  interfaces, was much more pronounced in the  $\beta$  annealed material than in the other conditions, and they suggested that the regular development of secondary cracks retards the overall crack rate.

Crack propagation is also influenced by the oxygen content of an alloy<sup>31,47-49</sup>. Two investigations showed that lowering the oxygen content from commercial purity levels (0.17-0.2 wt %) to levels of 0.08-0.12 wt % halved the crack rates over the range 0.01-1  $\mu\text{m}/\text{cycle}$  for Ti-6Al-4V<sup>47</sup> and Ti-6Al-6V-2Sn<sup>49</sup> alloys in the recrystallization annealed condition. However, very low oxygen contents result in significant losses in yield strength. Thus, an optimum combination of strength and crack propagation resistance might be obtained by choosing an alloy of intermediate oxygen content with a very resistant microstructure, such as the  $\beta$  annealed condition (although it is not known if oxygen content exerts a significant effect on crack rates in  $\beta$  annealed material<sup>47</sup>).

The influence of texture on crack propagation in air cannot be neglected. A strong texture can result in variations in growth rate, for different testing directions, by up to an order of magnitude<sup>50</sup> in the range  $5 \times 10^{-3}$ -5  $\mu\text{m}/\text{cycle}$ , although the variation appears to be typically a factor of 2-3<sup>50,51</sup>.

Of great importance is the coupling of environmental, textural and microstructural effects. In an aggressive environment the hexagonal  $\alpha$  phase cleaves on or near the basal plane<sup>45</sup>, and a texture lining up the basal planes in the crack plane should give greatly enhanced crack rates. That this is so is shown in figure 10. Strongly textured Ti-6Al-4V sheet with basal planes in the T-L crack plane was tested in air and 3.5 % aqueous NaCl<sup>52</sup>. The crack rates in air were within a factor of 2, but in 3.5 % aqueous NaCl a 20-fold acceleration was observed at low stress intensities. It is noteworthy that the frequency was high (50 Hz). Dropping the frequency to the gust range would significantly increase the crack rate differences in salt water, even up to much higher  $\Delta K$  values and crack rates<sup>45,53</sup>.

### Steels

Figure 11 is a summary plot of crack rates in air for many high strength steels<sup>28,54-72</sup>. The data for low alloy steels with yield strengths up to 1550  $\text{MN}/\text{m}^2$  and maraging and trip steels with yield strengths up to 2100  $\text{MN}/\text{m}^2$  all fall in a fairly narrow band (crack rates within a factor of 7). This is

a graphic illustration of the importance of Young's modulus, since the steels were of widely different compositions and microstructures.

Figure 11 also shows that in the ultrahigh strength range ( $\sigma_y > 1600 \text{ MN/m}^2$ ) the maraging and trip steels are superior to commercial purity low alloy steels and AM 355CRT and PH 15-7Mo. One reason for the inferiority of the low alloy steels appears to be susceptibility to brittle fracture at prior austenite grain boundaries<sup>63,73</sup>. The contribution of this fracture mode is lessened by reducing the content of elements which promote temper embrittlement (S,P,As,Sb,Sn)<sup>63</sup>, with a resulting improvement in fatigue crack growth resistance (figure 11, the data for high purity  $1\frac{1}{2}$  Ni-Cr-Mo steel). Second phase particles, however, have little or no effect<sup>66,71</sup>.

Many results<sup>62,64,74-76</sup> show that aggressive environments can greatly enhance crack propagation rates in high strength steels, the degree of enhancement being strongly dependent on test frequency.

Fatigue crack propagation resistance is not normally considered when selecting high strength steels for application. A notable exception is 10 Ni steel ( $\sigma_y \sim 1300 \text{ MN/m}^2$ ), which was chosen for U.S. Navy and Air Force evaluation programmes, the latter pertaining to the B-1 wing box. This steel has excellent weldability, fracture toughness and stress corrosion resistance<sup>70,72</sup>, and its resistance to corrosion fatigue is reported to be very good, as can be seen from figure 12: even at low frequencies the maximum enhancement in crack rates by water is only a factor of 2.

#### FRACTURE TOUGHNESS

It is generally found that fracture toughness decreases with increasing yield strength, figure 13. HAHN and ROSENFELD<sup>79</sup> explain this trend in the following way. They suggest that the plastic instabilities leading to fracture are a consequence of slip-induced breakdown of the submicron particles which strengthen the alloys, and that this breakdown occurs more rapidly at higher strength levels, where the particles are smaller and more fragile.

The preceding hypothesis remains unproven, but what is certain is that the amounts, distribution and morphology of alloy phases and the presence of coarse second phase particles have a large influence on toughness.

#### Aluminium alloys

Reducing the volume fraction of coarse ( $> 1 \mu\text{m}$ ) intermetallic particles, which are nuclei for large voids, can significantly improve the fracture toughness with little, if any, loss in strength<sup>5,38,79-81</sup>. In particular, limiting the Fe and Si contents is beneficial for 7000 series alloys, e.g. figure 14.

A much more difficult problem is posed by finer particles. These have a less strong but still important influence on toughness. Some of their constituent elements (e.g. Cu), while they lower toughness by aiding inclusion formation, on the other hand contribute to alloy strengthening by the age-hardening precipitation reaction.

Particles in the 0.03-0.5  $\mu\text{m}$  range appear to have dual, but contradictory, influences on toughness<sup>5</sup>. They suppress recrystallization or limit grain growth, thereby favouring transgranular, high energy fracture; but they also act as nuclei for sheets of small voids between the large voids nucleated at big inclusions<sup>5</sup>.

The finest particles ( $\leq 0.01 \mu\text{m}$ ) are the age-hardening precipitates. Their size is very sensitive to heat treatment, and they exert a major influence on toughness in 7000 series alloys: coarser and more numerous grain boundary particles promote intergranular fracture and low toughness<sup>5,82-84</sup>. This effect results in an influence of grain size and shape on toughness in conventionally processed and overaged materials, figure 15. Some data indicate that the matrix precipitates<sup>84</sup> and the width of the grain boundary precipitate free zone<sup>83,84</sup> have no influence on toughness. VAN LEEUWEN<sup>85</sup> found that a Structural Coarseness Index (a weighted average of the sizes of matrix and grain boundary precipitates and the widths of precipitate free zones at grain boundaries and larger particles) correlated moderately well with fracture toughness, figure 16.

Thermomechanical treatment (TMT) of 7000 series alloys<sup>11,86-89</sup> can achieve about 20 % improvement either in strength with no loss in toughness, or in toughness at a given strength level. Intermediate (ITMT) processing improves toughness by refining the grain size and separating coarse inclusions from the location of the grain boundaries<sup>88</sup>. Final (FTMT) processing exerts a strengthening by developing precipitate-stabilised, dense, uniform dislocations<sup>87,90</sup>. However, there are cost and control problems<sup>5</sup>, and the lack of improvement in notched fatigue strength<sup>11</sup> is discouraging. KAUFMAN<sup>5</sup> suggests that more significant strength-toughness improvements may come from powder metallurgy alloys; but again there is a cost problem.

#### Titanium alloys

Titanium alloys are inherently clean. The fracture toughness in commercial, high strength alloys is chiefly influenced by alloy phase distribution and morphology; texture; and interstitial oxygen and hydrogen contents.



Numerous investigations<sup>91-99</sup> have shown that the fracture toughness of an alloy is generally improved by increasing the amount of acicular  $\alpha$ , which is obtained by processing and/or heat treatment either in the  $\beta$  phase field or just below the  $\beta$  transus. However, as figure 17 shows, there is no overall superiority of  $\beta$  processed or heat treated alloys. In fact, the fracture toughness is consistently raised above the median values for all data only over the yield strength range 1000-1150 MN/m<sup>2</sup>. This result is no longer very surprising, since more recent work shows that besides the volume fraction of acicular  $\alpha$  (and the textural and interstitial content effects) the fracture toughness is, or may be, influenced by the volume fractions of primary  $\alpha$ , untransformed  $\beta$  and aged  $\beta$ ; by the volume fraction of plate  $\alpha$  precipitated at prior  $\beta$  grain boundaries; by the acicular  $\alpha$  plate thickness, length and spacing; by the width of the interface between  $\alpha$  plates and untransformed  $\beta$ ; and by recrystallization of primary  $\alpha$ .<sup>47,49,97,101-104</sup>

PATON et al<sup>47</sup> propose the following dependence of toughness on microstructure in  $\alpha$ - $\beta$  alloys:  $\beta$  anneal treatments are preferable to ( $\alpha$ + $\beta$ ) treatments; in ( $\alpha$ + $\beta$ ) processed and heat treated materials optimum toughness is achieved by recrystallization of  $\alpha$  and minimizing the volume fraction of untransformed or aged  $\beta$ ; heat treatments producing layers of  $\alpha$  at prior  $\beta$  grain boundaries can increase toughness.

Strong textures result in toughness varying by up to a factor of 2 for different crack plane orientations and fracture directions<sup>50,105</sup>. High toughness occurs when the crack plane and fracture direction most readily permit plastic flow<sup>22,106</sup>.

Lowering oxygen content from normal commercial levels ( $\sim 0.2$  wt.%) to extra low interstitial (ELI) levels (0.10-0.13 wt.%) almost doubles the fracture toughness<sup>107,108</sup> with about 10 % loss in strength. Even higher toughness occurs in very low oxygen (VLO) materials ( $< 0.08$  wt.%)<sup>109</sup>, but the yield strength loss is probably unacceptable for aircraft applications, and it is very difficult to routinely process titanium to less than 0.1 wt % oxygen.

Lowering the hydrogen content from 50 ppm to 10 ppm in Ti-6Al-4V by vacuum annealing raises the toughness by 50-100 %, <sup>110,111</sup> without loss in strength. At least for Ti-6Al-4V, it appears worthwhile to institute a vacuum treatment at a temperature selected to optimise the microstructure at the same time.

## Steels

Maraging steels usually have much higher fracture toughness than quenched and tempered steels at a given strength level, figure 13. COX and LOW<sup>112</sup> argue that the prime reason for this difference is that in low alloy steels large voids nucleate at inclusions (e.g. MnS) and link up via sheets of small voids nucleated at the carbide precipitates which strengthen the alloys; but in maraging steels only the large voids occur, at Ti(C,N) inclusions, and these have to grow until they coalesce.

Recent work<sup>113</sup> on D6ac steel confirms the primary role of carbide precipitates. However, other microstructural features are very important. Figure 18 shows the effect of unconventional heat treatments on the fracture toughness of 4340 steel: the toughness-yield strength trend has been shifted to that of maraging steels. This improvement was attributed to small amounts of retained austenite in a network surrounding the martensite plates and laths, and to the absence of twinned martensite plates<sup>114</sup>. A similar beneficial influence of retained austenite occurs in maraging steel<sup>69</sup>.

Qualitative effects of microstructural features on toughness in quenched and tempered steels are listed in Table II. At the present time it is not possible to establish the combined roles of a number of microconstituents<sup>114</sup>. Research proceeds on an experimental basis, sometimes with very unexpected results<sup>114</sup>.

Despite the importance of control of alloy phases, the fracture toughness of high strength steels can be most readily improved by greater microstructural cleanliness,<sup>79,112,115</sup> since inclusions provide easy nucleation sites for large voids<sup>112</sup>; hence the current trend to vacuum induction melting during consolidation<sup>70</sup>.

Figure 19 illustrates the dependence of toughness of several steels on volume fraction of weak second phases (inclusions). Toughness decreases with increasing volume fraction of inclusions, although the trend is less strong at higher strength levels. Limitations to figure 19 are that the data are few, and all second phases have been assumed equally detrimental<sup>79</sup>. The trends, particularly for higher strength steels, need to be refined, especially in the low volume fraction regime. Furthermore, not all second phases are equally detrimental. For example, MnS and Ti(C,N) have dominating influences in 4340 steel and maraging steels, respectively.<sup>112,116</sup>

As with fatigue crack propagation, fracture toughness is rarely a design criterion for high strength steels. Most components are safe-life, since critical defect sizes are generally too small to permit guaranteed in-service detection. The 10 M steel being evaluated by the U.S. Air Force and Navy has a plane strain fracture toughness in excess of  $200 \text{ MN/m}^{3/2}$ , which is sufficient to ensure fail-safety in thicknesses up to 4 cm at a yield strength of  $1300 \text{ MN/m}^2$ .<sup>70</sup>



## STRESS CORROSION

The trend of increasing stress corrosion susceptibility with increasing strength is a severe limitation on high strength materials. This is especially true of 7000 series aluminium alloys, and steels with yield strengths  $>1450 \text{ MN/m}^2$ , which tend to be highly susceptible to aqueous stress corrosion in both smooth and notched conditions. Nearly all titanium alloys are immune to aqueous stress corrosion in the absence of a crack. This helps to explain their exemplary performance in safe-life items. However, laboratory tests have shown that for many titanium alloys the propagation of a pre-existing crack is greatly accelerated by aqueous stress corrosion, so that this phenomenon must be considered in designing for fail-safety.

### Aluminium alloys

The main technique for controlling stress corrosion is artificial ageing (overageing in 7000 series alloys). Alloy composition and grain morphology are usually less important<sup>118</sup>. Thermomechanical treatments (TMT) indicate the possibility of avoiding the strength reductions which occur in conventionally processed and overaged materials.

Table III summarises salt water stress corrosion resistance of some conventionally processed 2000 and 7000 series alloys loaded in the short transverse direction. In the naturally aged T3 and T4 tempers the 2000 series alloys are highly susceptible. 2014 is also susceptible in the artificially aged T6 temper. All high strength 7000 series alloys are susceptible in the T6 temper, and the only effective means of improving the short transverse stress corrosion resistance is artificial ageing<sup>119</sup>. The disadvantage is a 10-15 % loss in yield strength.

Various theories, generally regarded as conflicting, relate stress corrosion susceptibility to the matrix precipitates<sup>120</sup>, to the precipitate free zone width<sup>121-123</sup>, to the grain boundary particle size<sup>124</sup>, and to the solute content of the precipitate free zone<sup>125</sup>. All these quantities tend to change simultaneously with heat treatment, thereby complicating the interpretation of results. VAN LEEUWEN<sup>85</sup> has considered this problem, using the Structural Coarseness Index (see previous section). He found that for 7079 alloy the Index gave a better correlation than any single quantity, but that for DTD 5024 there was a better correlation with the matrix precipitates. This result epitomises the state-of-the-art at the present time. It is noteworthy that VAN LEEUWEN et al<sup>126,127</sup> and THOMPSON and BERNSTEIN<sup>118</sup> advocate combination of the theories, with hydrogen embrittlement as a possible linking factor.

Both 2000 and 7000 series alloys benefit from thermomechanical treatment (TMT). Final (FTMT) processing enables 2024 to achieve a 16 % increase in yield strength over the T851 condition with equivalent stress corrosion resistance<sup>128</sup>, and for 7000 series alloys strength levels equivalent to T6 tempers can be obtained with T73 stress corrosion resistance.<sup>11,87,128,129</sup> These improvements were attributed to a uniform high density of dislocations and distribution of precipitates, the latter being comparable in size to those produced by overageing<sup>128</sup>. Intermediate (ITMT) processing of 7000 series alloys raises  $K_{I_{sc}}$  by 20-25 % in the short transverse direction, with about 5 % loss in longitudinal and transverse yield strengths<sup>88</sup>. The  $K_{I_{sc}}$  improvement is probably mainly due to avoidance of the elongated grain boundaries characteristic of the short transverse direction in conventionally processed material, since such boundaries provide an easy fracture path<sup>118</sup>.

Alloy composition has a modest effect on stress corrosion resistance<sup>118</sup>. In general, increasing the amount of major alloy additions in solid solution increases susceptibility<sup>119</sup>. Minor additions have two kinds of effect: elements like Cr, Mn and Zr tend to form intermetallics which pin grain boundaries and stabilise the wrought grain shape, thereby influencing stress corrosion resistance; Cr and Mn, and other elements like Cu, Li and Ag, influence stress corrosion through altering the ageing kinetics and sometimes the particle morphology<sup>118</sup>.

A final comment relates to data presentation. KAUFMAN<sup>5</sup> points out that aluminium alloy  $K_{I_{sc}}$  values do not provide safe levels of stress for all crack sizes. As an example, figure 20 shows clearly that the small flaw part of the  $K_{I_{sc}}$  analysis is unconservative, such that gross section stresses should be used.

### Titanium alloys

Metallurgical variables which influence the stress corrosion susceptibility of titanium alloys are the alloy composition and interstitial content, the slip character, and the structure, morphology and grain size of the alloy phases. These variables are somewhat interrelated, but their relative importance in determining susceptibility is different for the different types of alloy. Thus, for  $\alpha$  alloys the most important influence is composition; for  $\alpha$ - $\beta$  alloys it is phase morphology; and for  $\beta$  alloys it is phase structure, i.e. whether the microstructure is all- $\beta$  or aged to contain more than one phase.<sup>130</sup>

High strength commercial alloys (yield strengths  $> 850 \text{ MN/m}^2$ ) are mainly of the  $\alpha$ - $\beta$  type<sup>100</sup>. For these materials the principal determinants of susceptibility are the grain size, volume fraction and mean free path of the

susceptible  $\alpha$  phase<sup>130</sup>. Decreasing these parameters reduces susceptibility.<sup>130-135</sup> This is done by temperature cycling through the  $\beta$  transus during processing or heat-treatment: equiaxed  $\alpha$  transforms to fine plates, the volume fraction of  $\beta$  is increased, and the dispersion of  $\beta$  (which is immune and acts as a crack arrester<sup>130,131,133,134,136-138</sup>) between the  $\alpha$  plates reduces the mean free path. Figure 21 shows that  $\beta$  processing and/or heat treatment generally raises  $K_{I_{SCC}}$  above median values.

In  $\alpha$ - $\beta$  alloys interstitial content has less influence on stress corrosion susceptibility than phase morphology. Reduction in oxygen content to extra low interstitial (ELI) levels is beneficial mainly to toughness (see previous section), except when the total (Al+O) content is critical in determining susceptibility<sup>108</sup>. Similarly, reduction in hydrogen content (below 30 ppm) has little influence<sup>139</sup> except in otherwise highly susceptible alloys, when  $K_{I_{SCC}}$  can be doubled<sup>140</sup>.

Except in  $\beta$  alloys the mechanism of stress corrosion cracking involves cleavage of the  $\alpha$  phase on a plane  $13^\circ$ - $17^\circ$  from basal<sup>45,137,141</sup>. This results in a pronounced effect of texture in determining susceptibility in ( $\alpha$ + $\beta$ ) processed and heat treated materials<sup>136</sup>.  $\beta$  processing and/or heat treatment has the advantage of randomizing texture as well as providing resistant microstructures.

### Steels

The most important factor governing stress corrosion susceptibility in high strength steels is the strength level<sup>142</sup>. Figure 22 shows failure times for smooth specimens exposed to a marine atmosphere, and  $K_{I_{SCC}}$  values in salt water for a number of classes of steel. All steels with yield strengths above  $1450 \text{ MN/m}^2$  are susceptible to stress corrosion and tend to have low  $K_{I_{SCC}}$  values, irrespective of the type of alloy, composition, structure, or heat treatment.

Crack initiation in smooth specimens commences from surface pits,<sup>142,143,148</sup> which are prone to form at inclusions<sup>151-154</sup>. The most resistant materials are the martensitic stainless steels (e.g. 17-4 PH, 15-5 PH, PH 13-8Mo, AM 362) followed by semi-austenitic stainless steels (17-7 PH, PH 15-7Mo, AM 355).<sup>142,148</sup> The low alloy steels require protection, which is generally a combination of cadmium plating and overlying paint.

Crack propagation is mainly intergranular, along prior austenite grain boundaries.<sup>142,143,148,155-159</sup> Crack rates are generally faster at higher strength levels<sup>145,146</sup>, and are greatly dependent on microstructure. At a

given strength level the fastest cracking is observed in low alloy steels. Increasing alloy content lowers the crack velocity<sup>147,149</sup>, and maraging steels have crack rates  $\sim 10$  times less than low alloy steels<sup>146</sup>. Reducing the grain size of 4340 steel decreased the crack velocity<sup>160</sup>. CARTER<sup>146</sup> showed that crack rates in a 350 grade maraging steel could be varied over two orders of magnitude by changing the ageing temperature.

Information about microstructural influences on  $K_{I_{SCC}}$  is limited. Studies on 4340-type steels showed that there was no effect of prior austenite grain size<sup>143,160</sup>, and that the only quantities significantly altering  $K_{I_{SCC}}$  were the C and Mn contents<sup>143</sup>. CARTER<sup>146</sup> found no influence of heat treatment on  $K_{I_{SCC}}$  in a 350 grade maraging steel. For precipitation hardening stainless steels the generally lower  $K_{I_{SCC}}$  values in semi-austenitic alloys have been tentatively attributed to the presence of  $\delta$ -ferrite with an associated heavy concentration of carbide precipitates.

Since it is now reasonably established that the mechanism of stress corrosion cracking in high strength steels involves hydrogen embrittlement<sup>143</sup>, some work<sup>161,162</sup> on the susceptibility of different microstructures to hydrogen embrittlement is relevant here, and is summarised in Table IV.

There is a wide variation in crack velocities<sup>147</sup> and pitting resistance<sup>148</sup> for steels with similar  $K_{I_{SCC}}$  values. Thus, the susceptibility of a steel should not be rated solely from smooth specimen time-to-fail tests. On the other hand, the usefulness of a fail-safe fracture mechanics approach is limited to steels with yield strengths less than about  $1450 \text{ MN/m}^2$ , e.g. 10 Ni (figure 22b). This is because in the ultrahigh strength regime the  $K_{I_{SCC}}$  and  $K_{I_C}$  values are so low, and the stress corrosion cracking rates sufficiently high<sup>147</sup>, that no cracking can be tolerated during the life of a component.



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TABLE I

HIGH-CYCLE FATIGUE STRENGTHS OF HIGH STRENGTH MATERIALS <sup>2,3</sup>

MATERIALS	FATIGUE STRENGTH/ULTIMATE TENSILE STRENGTH
Aluminium alloys	0.25 - 0.3
Titanium alloys	0.3 - 0.6
High strength steels (1200 MN/m <sup>2</sup> )	0.45
High strength steels (1800 MN/m <sup>2</sup> )	0.4
Maraging steels	0.35

TABLE II

QUALITATIVE INFLUENCE OF MICROSTRUCTURAL FEATURES ON FRACTURE TOUGHNESS  
IN QUENCHED AND TEMPERED STEELS <sup>114</sup>

BENEFICIAL INFLUENCE	DETRIMENTAL INFLUENCE
Retained austenite network	Large carbides, sulphides
Lower bainite	Free ferrite, as grains or as platelets in upper bainite
Autotempered martensite with no interlath carbides	Upper bainite
Tempered martensite with no interlath films of carbides	Tempered plate martensite
	Twinned plate martensite

TABLE III

STRESS CORROSION RESISTANCE OF SOME 2000 AND 7000 SERIES ALUMINIUM ALLOYS  
IN 3.5 % NaCl WITH STRESSES IN THE SHORT TRANSVERSE DIRECTION 70,119

ALLOY-TEMPER DESIGNATION	" $K_{I_{SCC}}$ " ( $\text{MN}/\text{m}^{3/2}$ )	SMOOTH SPECIMEN THRESHOLD STRESS ( $\text{MN}/\text{m}^2$ )
2024-T351	9	48
2024-T4		< 55
2024-T62		300
2024-T851		275
2219-T37	< 9	< 69
2219-T62		> 220
2219-T87	~28	> 280
2014-T451	< 9	
2014-T651	< 8	55
7075-T651	8	48
7075-T7651	< 22	170
7075-T7351	~23	> 300
7178-T651	8	48
7178-T7651	~19	170
7050-T7651X	~10	
7050-T73651	25	
7475-T7351		> 300

TABLE IV

SUSCEPTIBILITY OF STEEL MICROSTRUCTURES TO HYDROGEN EMBRITTLEMENT 161,162

TYPE OF STEEL	MICROSTRUCTURE	DEGREE OF SUSCEPTIBILITY
Medium carbon-low alloy	Twinned martensite	high
Low carbon and low-to-high alloy content	Granular bainites	low
	Fully or predominantly slipped(lath) martensite, remainder lower bainite	low
	Slipped martensite and twinned martensite	medium
	Slipped martensite, twinned martensite and bainite	medium
	Fully or predominantly lath bainites (upper or lower)	medium



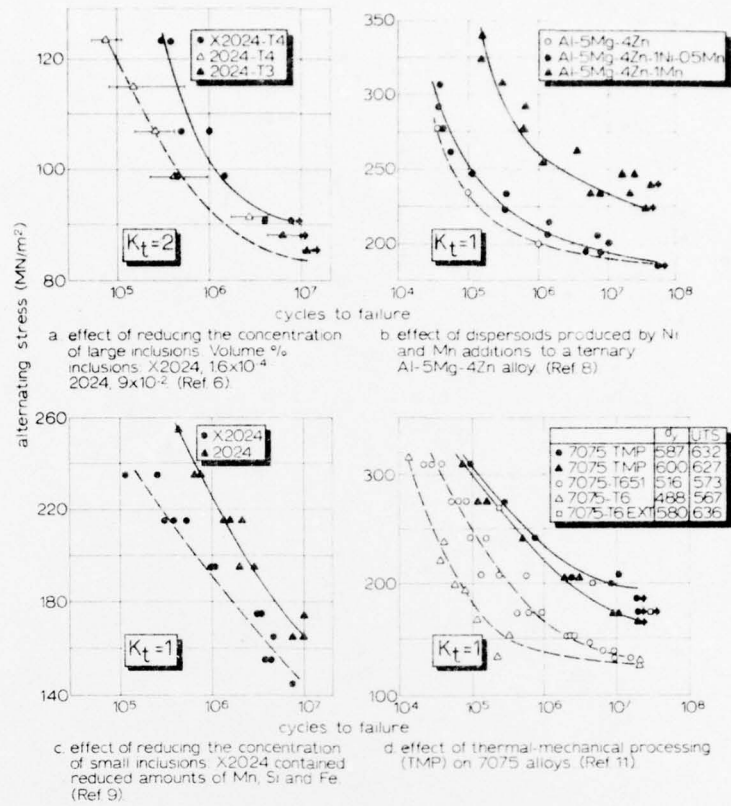


Fig 1. Microstructural effects on fatigue strength in aluminium alloys.

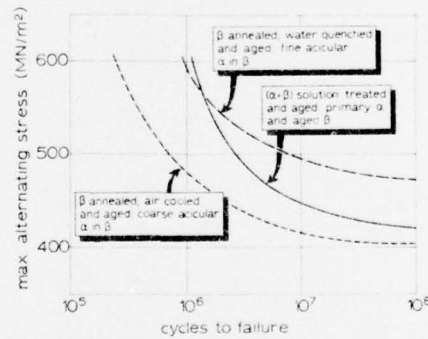


Fig 2 Effect of microstructure on fatigue life in Ti-6Al-4V forgings. All tests done with  $300 \text{ MN/m}^2$  mean stress (Ref 19).

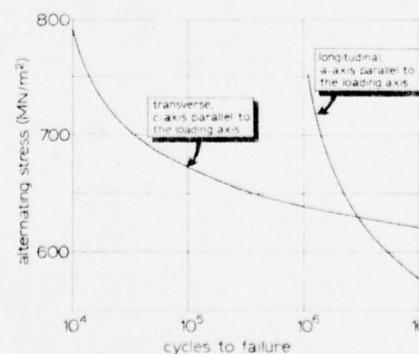


Fig 3 Effect of texture on fatigue life in Ti-4Al-4V plate. The stress ratio ( $R = S_{\min}/S_{\max}$ ) was 0.1 (Ref 24).

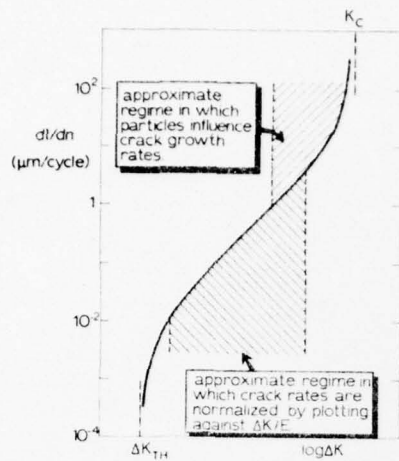


Fig 4 Schematic variation of fatigue crack growth rate,  $dl/dn$ , with stress intensity factor range,  $\Delta K$ .

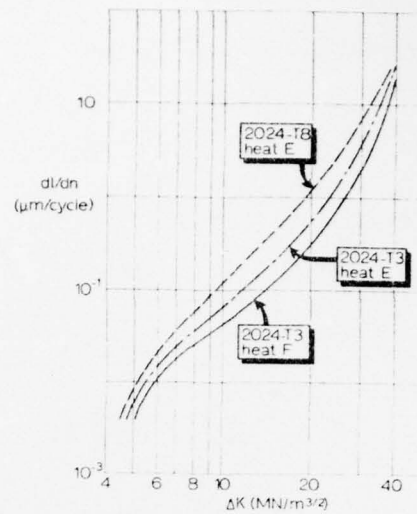


Fig 5 Influence of heat-to-heat variations and heat treatment on fatigue crack growth rates in 2024-T3 at  $R=0.037$  and 53Hz in normal air. (Ref 37).

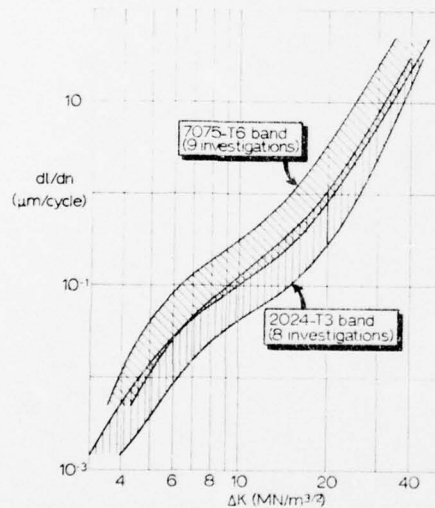


Fig 6. Comparison of fatigue crack growth rates in 2024-T3 and 7075-T6 in air of varying humidity with  $R=-1$  to 0.6 and frequencies of 20-133Hz. (Ref 35).

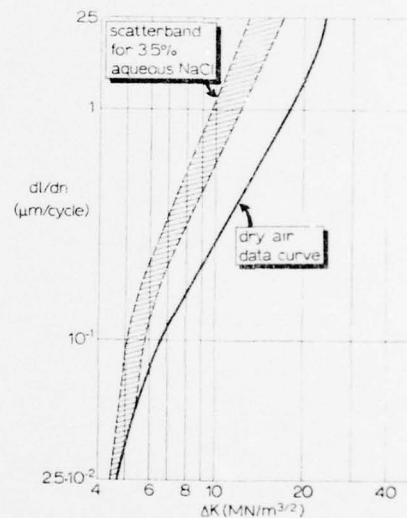


Fig 7 Comparison between crack growth rates in dry air and 35% aqueous NaCl for 7475 and 7050 alloys in various T7 tempers at  $R=0.1$  and 1Hz (Ref 38)

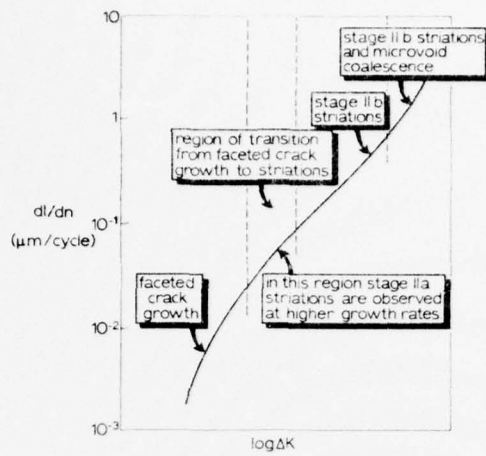


Fig 8. Schematic representation of fatigue crack propagation modes in titanium alloys in mild or nominally inert gaseous environments (Ref. 44).

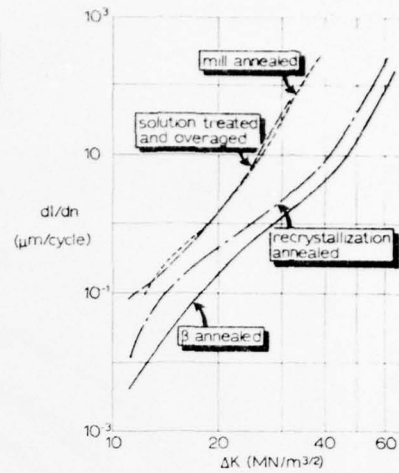


Fig 9. Crack growth rates in dry air for four microstructures of Ti-6Al-4V at  $R=0.1$  and 20 Hz (Ref. 47).

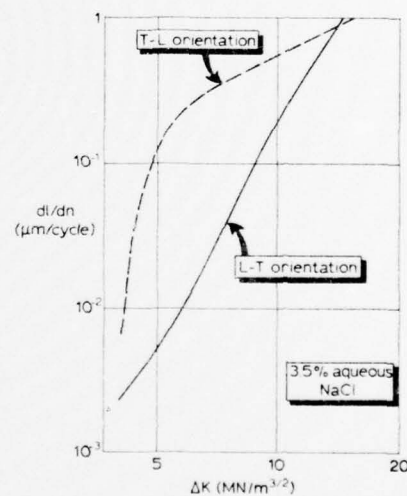
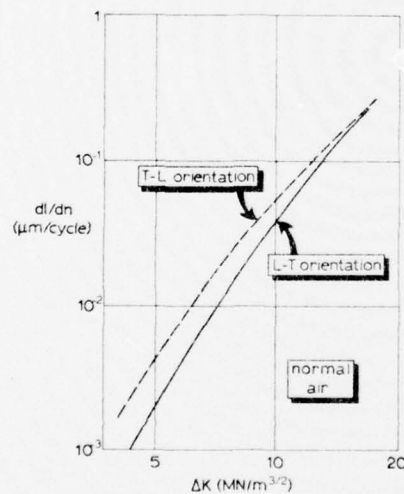


Fig 10. Crack propagation rates in air and 35% aqueous NaCl for strongly textured mill annealed Ti-6Al-4V sheet at  $R=0.667$  and 50 Hz (Ref. 52).

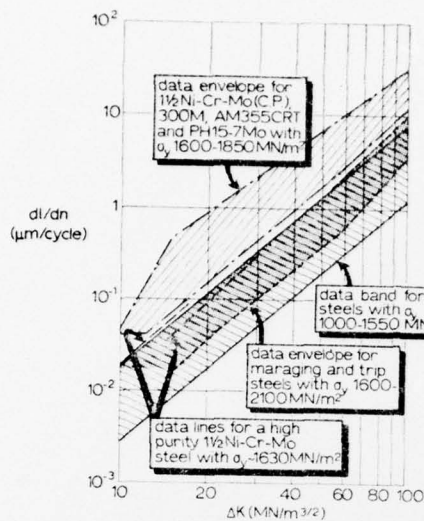


Fig.11. Fatigue crack growth rates in high strength steels in air at various frequencies and stress ratios (Refs. 28, 54-72).

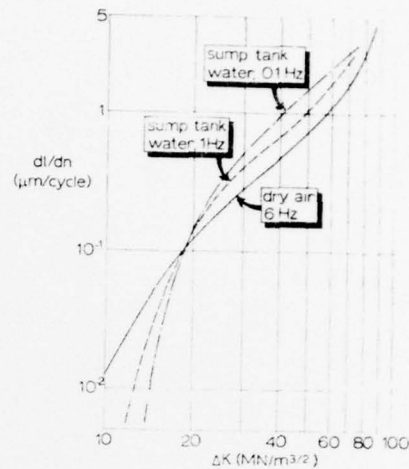


Fig.12 Crack growth rates in 10Ni steel at room temperature for various environments and frequencies with  $R=0.1$  (Refs. 70, 72).

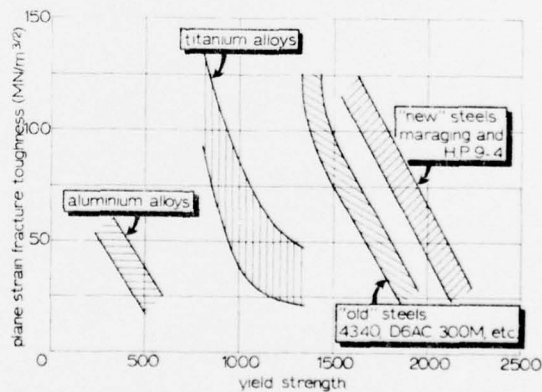


Fig.13. Fracture toughness as a function of yield strength for generic classes of high strength materials (Refs. 77, 78).



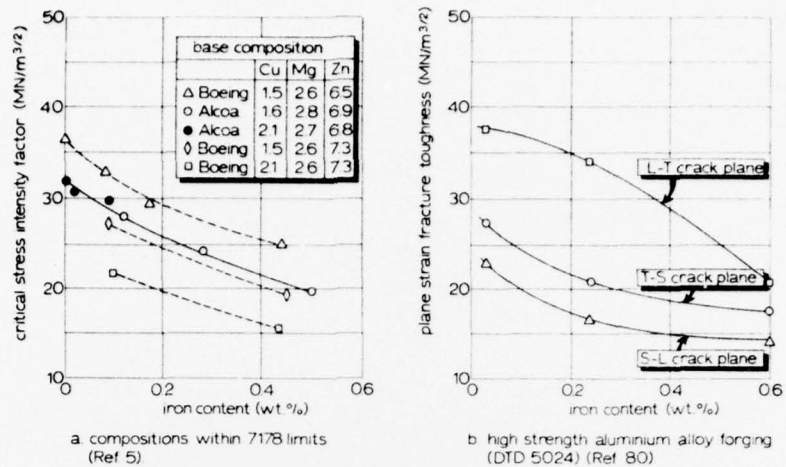


Fig 14. Effect of iron content on fracture toughness.

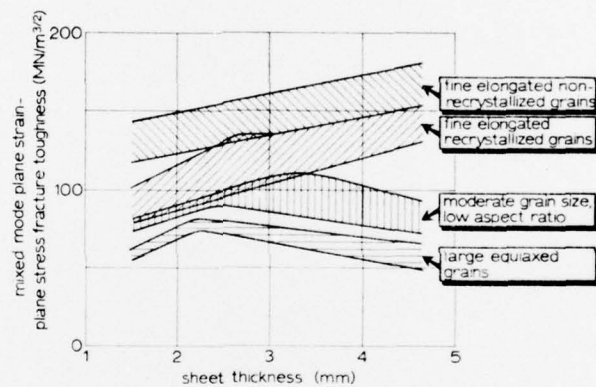


Fig. 15. Effect of grain size and shape on fracture toughness of overaged 7000 series aluminium alloy sheet. (Ref. 4).

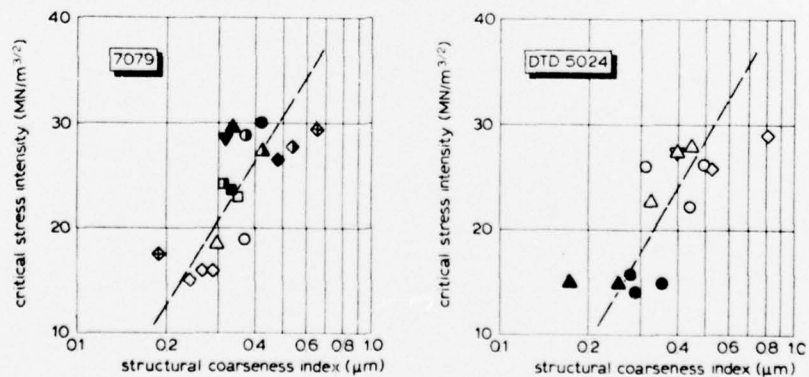


Fig 16. Fracture toughness of two aluminium alloys as a function of structural coarseness index. The symbols refer to various heat treatments (Ref 85).

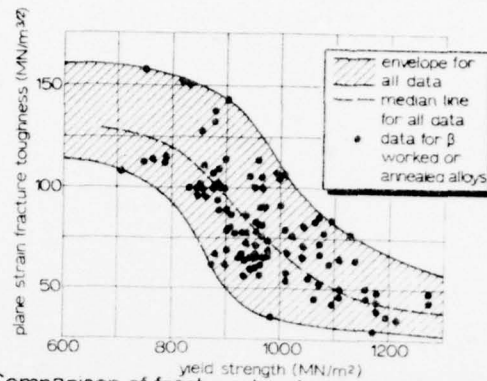


Fig 17. Comparison of fracture toughness envelope with data for  $\alpha$ , near- $\alpha$  and  $\alpha$ - $\beta$  titanium alloys  $\beta$  worked or annealed before further heat treatment (Ref 100).

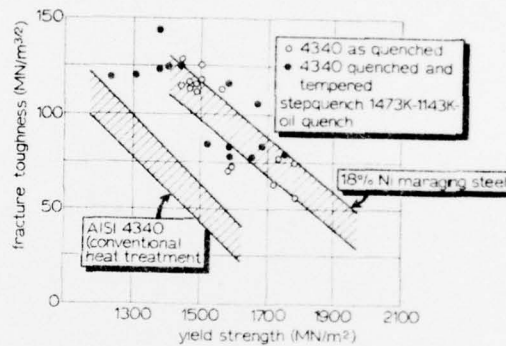


Fig 18. Fracture toughness vs yield strength diagram illustrating improvement of 4340 steel properties by experimental heat treatment (Ref 114).

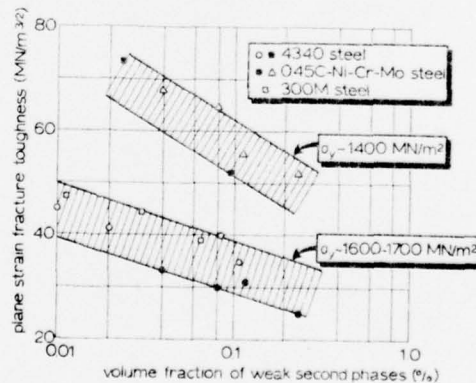


Fig 19. Fracture toughness vs volume fraction of weak second phases for some high strength steels. (Ref. 79).

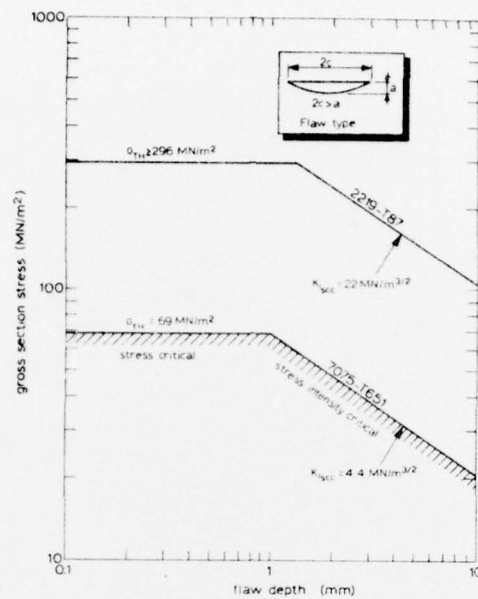


Fig. 20. Composite stress-stress intensity stress corrosion threshold characterization for two aluminium alloys exposed in a salt-dichromate-acetate solution (Ref 5).

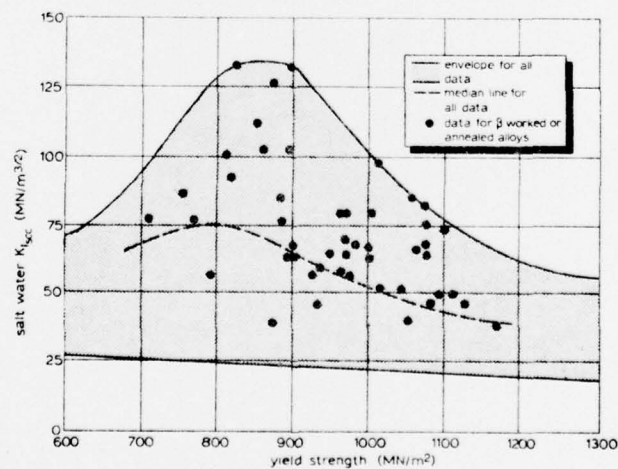


Fig 21. Comparison of  $K_{Isc}$  envelope with data for  $\alpha$ , near- $\alpha$  and  $\alpha$ - $\beta$  titanium alloys  $\beta$  worked or annealed before further heat treatment (Ref 100).

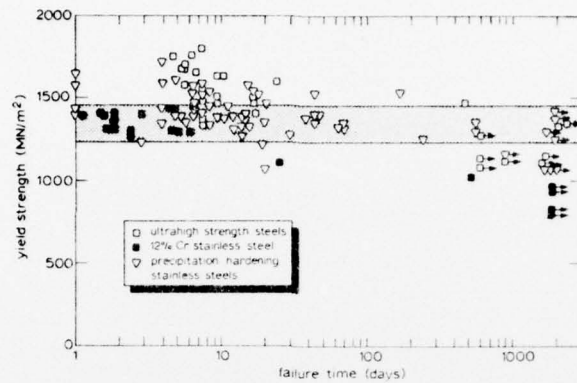


Fig 22a Stress corrosion failure times for high strength steels exposed to marine atmosphere at 75% of the yield strength. Each data point is the average lifetime of at least 5 specimens of a specific steel and heat treatment (Ref. 120).

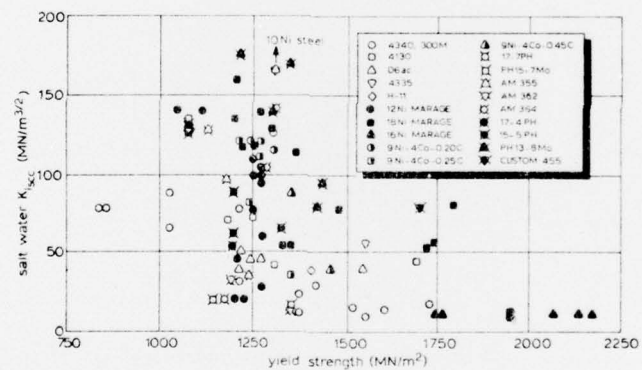


Fig 22b  $K_{Isc}$  data for a number of commercial steels tested in salt water (Refs 70, 117, 121-128).